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H Hassan
Case Western Reserve University

Alan K. Hellier
University of Wollongong, ahellier@uow.edu.au

Alan Crosky
University of New South Wales, a.crosky@unsw.edu.au

J Lewandowski
Case Western Reserve University

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Abstract

This paper reports fracture toughness data on aluminium alloy 6061 reinforced with 15 vol% of Al_2O_3 particulate of average size 13 µm. Composites were cast/extruded and heat-treated to give under-aged, peak-aged and over-aged samples, with both longitudinal (L-S) and transverse (T-S) properties being measured to determine the effects on toughness of the banded structure produced via extrusion. This anisotropy was reflected in the toughness results, primarily in the tearing modulus. Comparisons were made to the behaviour of layered/laminated composites where the effects are more dramatic and designed into the structure. This suggests that greater improvements to the fracture critical properties (at least in one orientation) could be obtained by purposely segregating the reinforcement during casting, then extruding/rolling to create a more segregated composite than that tested presently. The current results are compared with those from other similar composites reported in the literature.

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Fracture toughness of cast and extruded Al6061/15%Al_2O_3p metal matrix composites

H.A. Hassan[a,b], A.K. Hellier[c,*], A.G. Crosky[d] and J.J. Lewandowski[b]

[a]Department of Design and Production Engineering, Faculty of Engineering, Ain Shams University, Cairo 11571, Egypt; [b]Department of Materials Science and Engineering – 522 White Building, Case Western Reserve University, Cleveland, OH 44106, USA; [c]School of Mechanical, Materials, Mechatronic and Biomedical Engineering, University of Wollongong, Northfields Avenue, Wollongong, NSW 2522, Australia; [d]School of Materials Science and Engineering, UNSW Sydney (The University of New South Wales), Sydney, NSW 2052, Australia

* **CONTACT** A.K. Hellier, Tel: +61 (0)2 9349 8083; fax: +61 (0)2 9349 8083;
E-mail address: alanhellier@optusnet.com.au
H.A. Hassan, Tel: +20 2 2263 1619; E-mail address: Hala_Hassan@eng.asu.edu.eg
A.G. Crosky, Tel: +61 (0)2 9385 4424; E-mail address: a.crosky@unsw.edu.au
J.J. Lewandowski, Tel: +1 216 368 4234; E-mail address: jjl3@case.edu

HIGHLIGHTS

- We report J_{IC} fracture toughness and tearing modulus values for Al6061/15%Al_2O_3p.
- Cast/extruded and heat-treated composites were under-aged, peak-aged and over-aged.
- Both L-S and T-S properties were measured to investigate effects of banded structure.
- The tearing modulus results were significantly affected by the anisotropy.
- Our results are compared with those for similar composites reported in the literature.

ABSTRACT This paper reports fracture toughness data on aluminium alloy 6061 reinforced with 15 volume % of Al_2O_3 particulate of average size 13 μm. Composites were cast/extruded and heat-treated to give under-aged, peak-aged and over-aged samples, with both longitudinal (L-S) and transverse (T-S) properties being measured to determine the effects on toughness of the banded structure produced via extrusion. This anisotropy was reflected in the toughness results, primarily in the tearing modulus. Comparisons were made to the behaviour of layered/laminated composites where the effects are more dramatic and designed into the structure. This suggests that greater improvements to the fracture critical properties (at least in one orientation) could be obtained by purposely segregating the reinforcement during casting, then extruding/rolling to create a more segregated composite

than that tested presently. The current results are compared with those from other similar composites reported in the literature.

KEYWORDS Fracture toughness; J_{IC}; K_{JC}; Tearing modulus T; Aluminium alloy 6061; Alumina particulate; Metal-matrix composites (MMCs)

1. Introduction

Aluminium metal matrix composites (MMCs) have been of interest as engineering materials because of their higher specific stiffness and strength, as well as superior wear resistance, compared to unreinforced aluminium alloys. MMCs also possess high-temperature capability, high thermal conductivity and low coefficient of thermal expansion (CTE). Aluminium MMCs are produced by casting, powder metallurgy, in-situ development of reinforcements, and foil-and-fibre pressing techniques. Consistently high-quality products are now available in large quantities. They are applied in brake drums and rotors, clutch discs, pistons and other automotive components, as well as golf clubs, bicycles, machinery components, electronic substrates, extruded angles and channels, and a wide variety of other structural and electronic applications.

Particulate reinforced metal matrix composites (PRMMCs) are of special interest owing to the low cost of their raw materials and their ease of fabrication, making them suitable for applications requiring relatively high-volume production. The improved properties of PRMMCs result from the addition of hard ceramic particles (such as Al_2O_3 or SiC); the size, shape, volume fraction, interfacial properties and distribution of these (together with the fabrication method) determine the mechanical behaviour of the composite (Milan and Bowen 2004). However, PRMMCs have inferior ductility, toughness and low-cycle fatigue properties compared with unreinforced alloys, often limiting their usefulness in practice. Reinforcements of between 15 volume % and 25 volume % (but most commonly 20 volume %) have found commercial application, as these are considered to provide the optimal combination of structural properties (Hooker and Doorbar 2000). For example, Alcan has manufactured car brake discs from a (cast) A359 Al-Mg-Si alloy reinforced with 20 volume % SiC particles (Kaczmar et al. 2000). Also, more recently, the Chevrolet Corvette driveshaft has been made of aluminium 6061 reinforced with 20 volume % Al_2O_3 particles, which exhibits a 36% increase in specific modulus over steel (Chawla and Chawla 2006). The present work investigates the toughness and fracture behaviour of a cast and extruded aluminium 6061/15 volume % alumina microsphere (of average size 13 μm) PRMMC as a function of heat treatment and orientation. The current results are compared with those from other similar composites reported in the literature.

2. Experimental procedure

2.1. Materials

The first set of materials tested was a cast and extruded aluminium alloy 6061 as well as 6061 reinforced with 15% by volume of Al_2O_3 particulate of average size 13 μm (Klimowicz and Vecchio 1990, Liu and Lewandowski 1993b). The 6061 composite was originally processed by Dural Aluminum Composites Corporation (DACC), San Diego, CA via proprietary molten metal and mixing technology. Direct chill-casting techniques were used to create 18 cm diameter x 110 cm long billets, with details provided elsewhere (Klimowicz and Nguyen-Dinh 1989). Table 1 gives the chemical composition of the unreinforced aluminium matrix alloy. The cast billets were cut to 45 cm lengths followed by forward extrusion using a 3850 ton hydraulic press with a bore diameter of 21 cm. Billet temperatures were 343-357°C, with extrusion exit speeds of 13 m/min. The composite materials were received as 0.5 m long extrusions with 19 mm x 76 mm rectangular cross-section. A 3-D optical metallographic view of the composite is provided in Figure 1. L is the length direction; T is the transverse direction; and S is the short transverse direction. At this magnification, it is clear that the reinforcement is somewhat banded, with preferential alignment of the particulate along the extrusion direction. No residual casting porosity was detected after such extrusion, consistent with previous work (Klimowicz and Vecchio 1990, Liu and Lewandowski 1993b).

2.2. Heat treatment

The unreinforced material was solution treated at 510°C for 4 hours, water quenched (WQ) and then artificially aged at either 175°C for 2 hours, or at 175°C for 100 hours, to produce an under-aged (UA) and over-aged (OA) microstructure, respectively. The composite was heat treated to attain a variety of conditions. After solution heat treatment at 510°C for 4 hours and a water quench (WQ), the composite was aged at 175°C for the following times: 40 min (UA1), 2 hrs (UA2), 10 hrs (PA), 40 hrs (OA1), and 100 hrs (OA2). These heat treatments were chosen on the basis of standard (Rockwell F-Scale; 60 kg load, 1/16″ ball, 15 second dwell time) hardness measurements (American Society for Testing and Materials 2016) on the materials at various ageing times, as shown in Figure 2, and were selected to provide a range of matrix ageing conditions. Transmission electron microscopy of the UA and OA microstructures is provided elsewhere (Strangwood et al. 1990; Strangwood et al. 1991).

2.3. Specimen orientation

In order to evaluate the effects of specimen orientation on fracture properties, specimens were cut in two perpendicular directions as shown in Figure 3. The first orientation, where the specimen length is along the rolling direction, is termed L-S and the orientation perpendicular to this is termed T-S. The first letter denotes the direction perpendicular to the notch plane; the second letter denotes the anticipated (Mode I) direction of crack propagation. Comparison with the 3-D microstructure in Figure 1 reveals that the longitudinal specimens will sample the reinforcement in a different manner than will the transverse ones. The former geometry resembles a layered/laminated structure with alternating bands of unreinforced aluminium and highly clustered composite (i.e. reinforced) regions, in contrast to the latter orientation. In both cases the crack was grown along the short transverse direction.

2.4. Fracture properties

Fracture properties were evaluated in general accordance with the standard procedure for J testing, ASTM E813 (American Society for Testing and Materials 1981). The tests were conducted as J tests and converted to K_{JC} data in order to determine the tearing modulus (T) differences between the longitudinal and transverse data. The unreinforced alloy, as well as the composite, were tested using three-point-bend specimens, the dimensions of which were 12.7 mm x 12.7 mm x 50 mm, as shown in Figure 4. The initial notch in all cases was 2.5 mm deep; a fatigue pre-crack was started from this notch and grown in accordance with ASTM E813 (American Society for Testing and Materials 1981). The crack length was continuously monitored using an electrical potential drop system. The calibration curve for both monolithic and composite samples was based on previous work on similar specimens. Single tests were conducted to generate the J versus Δa plots shown in Figures 5 and 6 for longitudinal UA2 MMC and longitudinal OA2 MMC, respectively. Control UA and control OA J versus Δa fitted data are shown in Figures 7 and 8, respectively. Tables 2–4 summarize the individual results. A few interrupted tests were conducted in the L-S orientation in order to examine the crack tip regions in samples exhibiting a non-zero tearing modulus. This was not possible for the T-S samples due to the non-existent tearing modulus, as described later.

3. Results and discussion

3.1. Hardness

Hardness data depicted in Figure 2 are generally consistent with much other work reported in the literature (Christman and Suresh 1988; Dutta and Bourell 1990) that reveal an acceleration of the ageing response in such composites. Much of the previous work was conducted on powder metallurgy processed composites, while the present composites were prepared via casting and subsequent extrusion. In that regard, solute segregation to the reinforcement/matrix interfaces in the present composites is significantly different than that obtained in powder metallurgy composites, and particularly those that are consolidated below the solidus temperature, as quantified using TEM and summarized elsewhere (Strangwood et al. 1990; Strangwood et al. 1991). While significant segregation of Mg to the reinforcement/matrix interfaces is evident in the cast/extruded product, there remains some acceleration of the ageing response exhibited by the composite in comparison with the unreinforced material.

Others (Myriounis et al. 2008a; Myriounis et al. 2008b; Myriounis et al. 2008c; Myriounis et al. 2009) have found hardness increases near the interfaces. This was not explicitly tested for in the present work, due to the difficulty of placing micro-hardness indents near the interfaces while avoiding contributions from the adjacent and/or underlying reinforcement. Such experiments would be valuable to explore in future work, perhaps even using nano-indentation.

3.2. Fracture toughness

While the hardness data is generally consistent with the behaviour exhibited in precipitation-hardened monolithic alloys, as well as composites where hardness increases to a peak condition followed by some loss in hardness, the fracture toughness drops with an increase in hardness (i.e. strength) and does not recover on further ageing. Rather, the toughness of the composite during over ageing remains at nearly the same value attained during peak ageing, while that of the monolithic matrix exhibits some recovery of toughness, as indicated in Tables 2–4. These observations are consistent with those reported in the literature for a number of PRMMCs (Klimowicz and Vecchio 1990; Lewandowski et al. 1989; Manoharan and Lewandowski 1990) as reviewed previously (Hunt et al. 1993) and more recently (Hassan and Lewandowski 2018).

In various powder metallurgy PRMMCs (Lewandowski et al. 1989; Manoharan and Lewandowski 1990; Singh and Lewandowski 1993) the effects of continued ageing past the peak-aged condition were to change the micro-mechanisms of failure from one that was dominated by cavitation via reinforcement fracture, and subsequent link-up through the matrix, to one where some void formation occurred via particulate fracture, with additional damage accumulation at particulate/matrix interfaces and in the matrix between particles (Klimowicz and Vecchio 1990; Singh and Lewandowski 1993). While such failure processes can be effectively suppressed via the superposition of hydrostatic pressure during testing (Lewandowski and Lowhaphandu 1998; Liu et al. 1989; Liu and Lewandowski 1993a, 1993b; Mahon et al. 1990; Singh and Lewandowski 1993; Vasudevan and Richmond 1989), they are accelerated in the presence of tensile triaxial stresses like those developed near a notch or fatigue pre-crack, as well as those developed near closely-spaced, non-deforming reinforcement particles (Bao et al. 1991; Christman et al. 1989). The tensile triaxial stresses developed during tensile straining of such PRMMCs contribute to the lower tensile ductility and toughness of these PRMMCs in comparison to their unreinforced matrices, while recovery of the ductility and/or toughness can only be obtained via the superposition of sufficient levels of compressive hydrostatic pressure, or via extrinsic toughening approaches, as discussed later.

In the present case, the fracture surfaces of all composite samples revealed subtle differences in fractography between the tougher UA samples and the samples that exhibited lower toughness with continued ageing. In particular, and somewhat similar to that reported by many others (Hunt et al 1993, Lewandowski et al 1989, Liu and Lewandowski 1993b; Manoharan and Lewandowski 1989, Singh and Lewandowski 1993) including experiments conducted on a material very similar to that tested presently (Klimowicz and Vecchio 1990), fracture in the UA materials appeared to nucleate via fracture of the reinforcement, followed by link-up through the matrix, creating a fracture surface with dimples centred on the fractured reinforcement particles. In contrast, the lower toughness samples that were obtained on continued ageing to peak strength and beyond exhibited a population of smaller dimples on the fracture surface between the voids centred on the fractured reinforcement, also similar to previous work (Klimowicz and Vecchio 1990) and that captured in a recent extensive review article (Hassan and Lewandowski 2018). The crack tip region in the L-S OA2 MMC, Figure 9, shows damage in the form of particulate fracture, damage at particulate/matrix interfaces, and extensive failure in the matrix between reinforcement, albeit with a positive value for the tearing modulus (i.e. $T = 0.9$) due to the layered nature of the L-S sample shown in Figure 1. Attempts at imaging the crack tip in the T-S samples in the OA conditions were unsuccessful due to the non-existent tearing modulus (i.e. $T = 0$) present in this orientation.

The behaviour of Al-Mg-Cu based composites reinforced with particulate alumina in the annealed condition has been investigated by Kamat et al. (1989). In that system, the fracture toughness, for reinforcement volume fraction 2-20% was found to increase slightly with reinforcement spacing, providing that the particle size was less than a critical value (i.e. 15 μm). Their stress intensity values at fracture, at least for a limited range of the small sizes of Al_2O_3 and lower volume fractions, were compatible with a Rice and Johnson (1970) type model. Interestingly, recent toughness results on nano-composite aluminium alloys (Hassan and Lewandowski 2008a) reveal toughnesses of similar magnitude to those reported previously, albeit in a system with much closer spacing of reinforcement due to the nano-sized particles. This apparently results from clusters of particles nucleating fracture to create one dimple in the nano-composite aluminium alloys, as reviewed elsewhere (Hassan and Lewandowski 2008a).

Initial attempts at rationalizing the present results have focused on a model first suggested by Rice and Johnson (1970). In such a model, all particles are considered to crack or decohere ahead of the major crack tip and at a low strain value. Experimental observations of the crack tip regions in these materials have been provided elsewhere (Manoharan and Lewandowski 1989; Manoharan and Lewandowski 1990). The region of intense plastic flow is limited to a volume of width δ, a value which could correspond to the interparticle/reinforcement spacing. If such a relation were to hold, the following equation applies:

$$\frac{J_{IC}}{\sigma_f} = \delta \tag{1}$$

The region of intense plastic flow can be approximated by the size of the dimple containing the fractured alumina particle. Analysis of the fracture surfaces reveals this distance in these composites to be about 15 μm. As the strengths for these composites are roughly 300 MPa for the ageing conditions investigated, the above relationship would predict a J_{IC} value of 4.5 kJ/m^2, in the range of those exhibited presently. The significantly lower toughness obtained in the more heavily-aged samples, along with fracture surface observations of a reduced dimple size between the reinforcement particles and the crack tip regions like that shown in Figure 9, suggests that a second population of void nucleating features between the reinforcements may also act as potent void nucleation sites, thereby further reducing the toughness.

As suggested above, the differences between the L-S and T-S toughness values primarily relate to the magnitude of the tearing modulus, *T*, summarized in Tables 3 and 4 respectively. The L-S samples consistently exhibited higher values for *T* in comparison to the T-S samples, although the L-S values for *T* in this composite were somewhat less than that reported for the powder metallurgy PRMMCs discussed earlier. In the present work, this anisotropy in *T* values appears to relate to the anisotropic structure (i.e. banding of reinforcement) present in such cast and extruded materials, as shown previously in Figure 1.

In the L-S orientation, the banding of reinforcement creates reinforcement-rich and lean regions, and resembles that of a layered/laminated system tested in the crack arrestor orientation. Much previous work on layered/laminated composites as well as nano-composites (Ellis and Lewandowski 1991; Ellis and Lewandowski 1994; Hassan and Lewandowski 2008b; Leseur et al. 1996; Pandey et al. 2001; Wu et al. 1996) has shown significant improvements to the initiation and/or growth toughness via such extrinsic approaches. However, the properties of such layered systems are typically worse in the T-S orientation, as shown presently, due to the distribution of particulate leading to unstable fracture instead of interrupted crack growth via the reinforcement-lean regions. A *T* value of zero implies that a crack in the T-S orientation exhibits fully catastrophic behaviour.

There is a scarcity of fracture toughness data for different orientations in PRMMCs to compare with the present results. Park et al. (2008) have measured the short rod and short bar (chevron-notch) fracture toughnesses K_{IV} in three orientations for a 20 volume % microsphere Al_2O_3–Al 6061-T6 PRMMC, with particulate of average size 20 μm, in the form of an extruded 19 mm diameter rod manufactured by a molten metal mixing method. The average K_{IV} fracture toughness values in the R-L, C-R and L-R orientations were found to be 16.5, 19.1 and 20.2 MPa.m$^{1/2}$ respectively. Here the L-R and C-R orientations correspond to the L-S and T-S orientations in the present work. It can be seen that K_{IV} in the C-R orientation is slightly less than in the L-R orientation, while K_{JC} for the peak-aged condition in the T-S orientation is also slightly lower than in the L-S orientation (as shown in Tables 3 and 4). The magnitudes of these fracture toughness values are all very similar, as expected.

Song and Han (1997) have measured the static fracture toughness K_{IC} in the C-R orientation for a discontinuously reinforced aluminium 6061-T6 MMC containing 15 volume % of Al_2O_3 long fibres as 17.8 MPa.m$^{1/2}$. This MMC was fabricated by a squeeze infiltration method, then cast into an ingot prior to heat treatment. The value is close to the present peak-aged K_{JC} value of 19.0 MPa.m$^{1/2}$ measured in the T-S orientation, despite the considerable difference between the forms of the reinforcement in the two cases. The L-T toughness values across a range of aging conditions for a material very similar to that reported presently (Klimowicz and Vecchio 1990) are consistent with the values reported presently for the L-S samples, although no tearing modulus information was reported in that work.

4. Conclusions

The effects of ageing condition on the hardness, crack initiation and growth toughness of cast and extruded Al6061-15% Al_2O_3p composites were determined in the L-S and T-S orientations. While hardness values exhibited an increase with ageing and some slight reduction on over ageing, the toughness values decreased up to the peak ageing condition and did not significantly change past the peak, in contrast to the behaviour of the monolithic matrix that showed recovery of the toughness during ageing past the peak in hardness/strength. Fracture surface observations revealed dimpled fracture with voids centred on fractured reinforcement particles, while the size of dimples between the reinforcement decreased with continued ageing as shown in much other work. Direct examination of crack tip regions revealed a smaller crack opening displacement at failure for the OA2 material compared to the UA materials, also consistent with previous observations on other MMCs. Estimates of the toughness using a Rice and Johnson model provided reasonable agreement for the UA materials, with the assumption of void nucleation from each reinforcement particle.

Although the fracture initiation toughness was similar in the L-S and T-S orientations, anisotropy in tearing modulus, *T*, was evident and likely due to the banding of reinforcement produced during extrusion. A *T* value of zero implies that a crack in the T-S orientation exhibits fully brittle behaviour. Comparisons were made to the behaviour of layered/laminated composites where the effects are more dramatic and designed into the structure. This suggests that greater improvements to the fracture critical properties (at least in one orientation) could be obtained by purposely segregating the reinforcement during casting, then extruding/rolling to create a more segregated composite than that tested presently.

Disclosure statement

No potential conflict of interest was reported by the authors.

Notes on contributors

Hala Abd El-Hakim Hassan is a Professor in the Design and Production Engineering Department, Faculty of Engineering at Ain Shams University, Cairo, Egypt. She received BS and MS degrees from Ain Shams University plus a PhD from both Ain Shams and CWRU via the Channel System. She has received two Junior Scientist Development Visit Grants from the US-Egypt Joint Board on Scientific and Technological Cooperation, in addition to a Fulbright Scholar Award and the 2008 Egyptian Government Encouragement Prize in Advanced Technology and Science which serves the engineering field.

Alan K. Hellier, FIEAust, CPEng, NER, RPEQ, APEC Engineer, IntPE(Aus), is an Honorary Senior Fellow in the School of Mechanical, Materials, Mechatronic and Biomedical Engineering at Wollongong University, Australia. He graduated from Cambridge University with a BA and MA in Natural Sciences (Metallurgy and Materials Science) followed by a PhD in Metallurgy from UNSW Sydney. Postdoctoral periods were spent at Cornell University and University College London. Following this he was a Lecturer in the School of Materials Science and Engineering at UNSW Sydney and a Senior Research Scientist at the Australian Nuclear Science and Technology Organisation. He subsequently worked as a Visiting Fellow in the School of Mechanical and Manufacturing Engineering at UNSW Sydney. An MBA was recently gained from Southern Cross University, Australia. His research interests include fatigue and fracture of metals and MMCs, failure investigation, stress analysis and fatigue assessment of welded joints, as well as computer structural modelling. He has published nearly 80 journal and conference papers, with an h-index of 13.

Alan G. Crosky is a Professor in the School of Materials Science and Engineering at UNSW Sydney. His principal research interests are in the areas of failure analysis and composite materials. He is the recipient of several government and industry research grants totalling \$2.3m since 2003, and has published 180 journal and conference papers, with an h-index of 18. Prior to joining UNSW, he led the Materials and Process Engineering Group at Hawker de Havilland Ltd in Bankstown, Sydney. His work there involved materials and process engineering support to aircraft structures fabrication, airframe overhaul, as well as overhaul of heavy piston and light turbine aeroengines.

John J. Lewandowski is the Arthur P. Armington Professor of Engineering II at Case Western Reserve University as well as Director of the Advanced Manufacturing and Mechanical Reliability Center (AMMRC). He established the AMMRC in 1987. It currently houses unique equipment valued in excess of \$4.5m, and has been used by his research group (and others) in areas ranging from advanced aerospace materials to nuclear and biomedical systems. Research and teaching interests are primarily in the areas of processing/structure/property relationships in ferrous and non-ferrous engineering materials; effects of superposed pressure on deformation/fracture; fatigue and fracture of intermetallics, composites, bulk metallic glass, and layered/laminated materials; deformation processing; blast-resistant materials; advanced engineered materials systems; fracture and fatigue behaviour of biomedical materials; failure analysis; and additive manufacturing. Publications and presentations exceed 320 and 950, respectively, with an h-index > 55. His top 55 papers have been cited over 6,000 times with recent major reviews on Additive Manufacturing, Fracture and Fatigue of Materials used in Biomedical Applications, and Metal Matrix Composites. His recent Nature Materials (NM) paper on metallic glasses was selected for a

News and Views Article in NM, was also selected as an Editor's Choice Paper in Science, and was additionally selected by NM Editors as one of the twenty most influential papers published in NM from 2002-2012.

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Tables

Table 1. Chemical composition of unreinforced aluminium matrix alloy 6061.

Wt%	Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Al
6061	0.57	0.35	0.20	0.01	0.84	0.05	0.05	0.13	Bal.

Table 2. Control toughness data on unreinforced cast/extruded aluminium matrix alloy 6061.

Heat treatment	Fracture toughness, J_{IC} (kJ/m^2)	Tearing modulus, T
Solution treated 510°C/4 hrs/WQ + 175°C/2 hrs (UA)	19.5	2.9
Solution treated 510°C/4 hrs/WQ + 175°C/10 hrs (PA)	17.0	2.9
Solution treated 510°C/4 hrs/WQ + 175°C/100 hrs (OA)	18.5	2.85

WQ = Water quenched; UA = Under-aged; PA = Peak-aged; OA = Over-aged

Table 3. Longitudinal (L-S) fracture properties of cast/extruded Al6061/Al_2O_3/15% composites.

Heat treatment	J_{IC} (kJ/m^2)	K_{JC} (MPa.m$^{1/2}$)	Tearing modulus, T
Solution treated 510°C/4 hrs/WQ + 175°C/40 min (UA1)	4.6	21.5	0.9
Solution treated 510°C/4 hrs/WQ + 175°C/2 hrs (UA2)	4.2	20.5	0.9
Solution treated 510°C/4 hrs/WQ + 175°C/10 hrs (PA)	3.8	19.5	0.9
Solution treated 510°C/4 hrs/WQ + 175°C/40 hrs (OA1)	3.4	18.5	0.9
Solution treated 510°C/4 hrs/WQ + 175°C/100 hrs (OA2)	3.2	18.0	0.9

Table 4. Transverse (T-S) fracture properties of cast/extruded Al6061/Al_2O_3/15% composites.

Heat treatment	J_{IC} (kJ/m^2)	K_{JC} (MPa.m$^{1/2}$)	Tearing modulus, T
Solution treated 510°C/4 hrs/WQ + 175°C/40 min (UA1)	4.6	21.5	0.0
Solution treated 510°C/4 hrs/WQ + 175°C/2 hrs (UA2)	4.2	20.5	0.0
Solution treated 510°C/4 hrs/WQ + 175°C/10 hrs (PA)	3.6	19.0	0.0
Solution treated 510°C/4 hrs/WQ + 175°C/40 hrs (OA1)	3.2	18.0	0.0
Solution treated 510°C/4 hrs/WQ + 175°C/100 hrs (OA2)	3.2	18.0	0.0

Figures

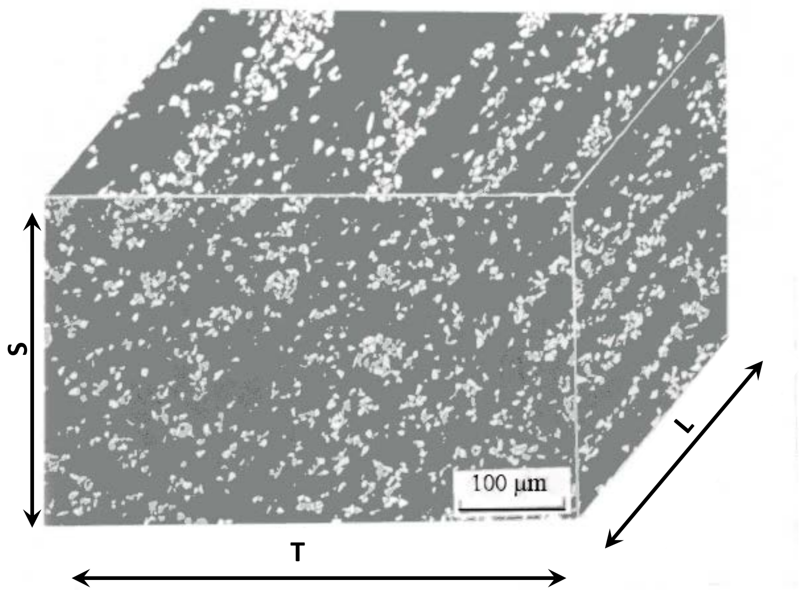

Figure 1. Three-dimensional view of composite.

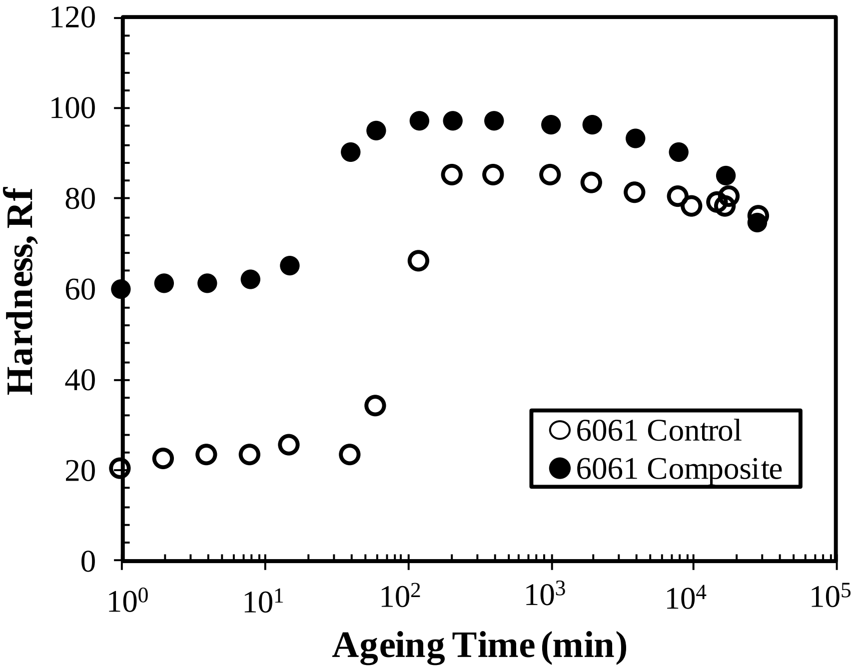

Figure 2. Ageing hardness curves for 6061 control and 6061 composite.

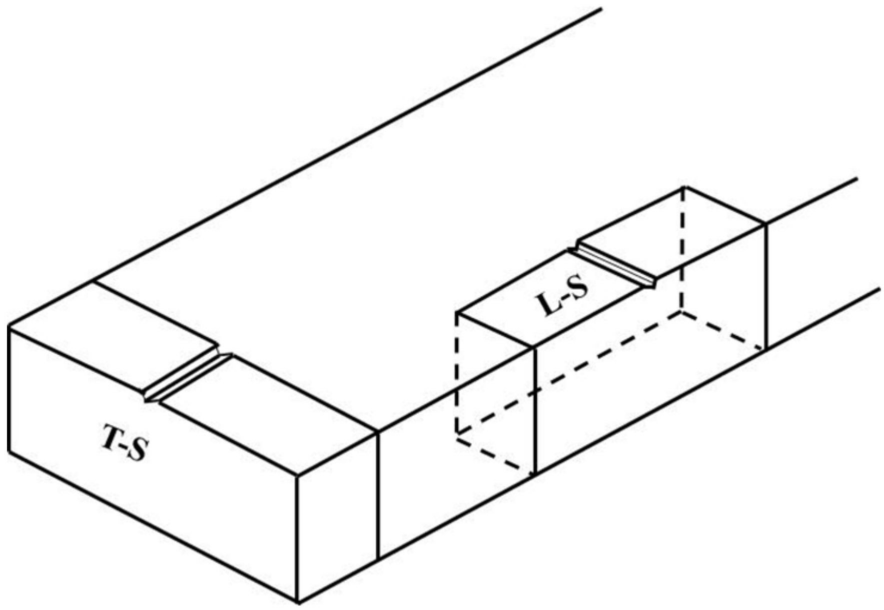

Figure 3. Orientations of specimens cut from plate.

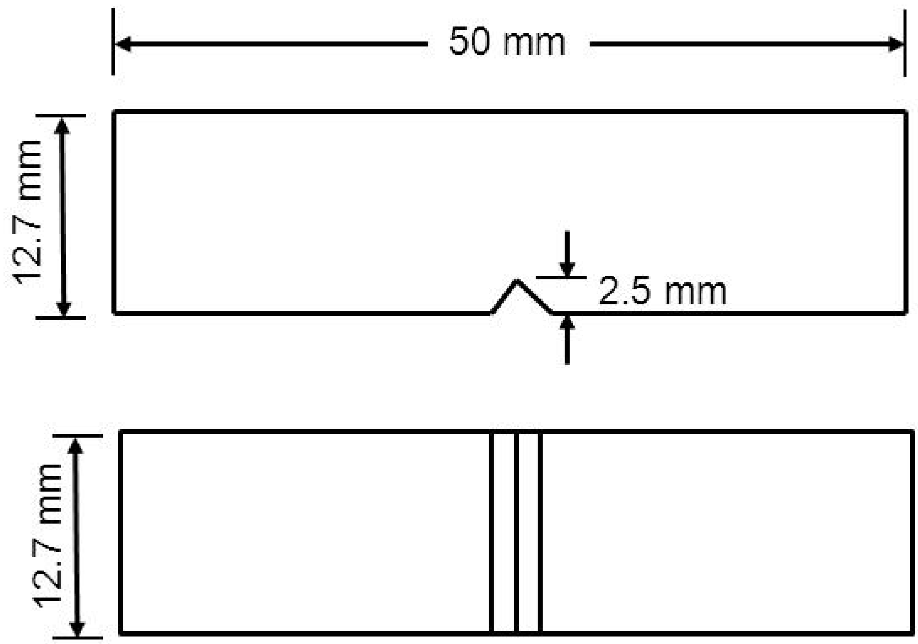

Figure 4. Three-point-bend specimen used for fracture toughness testing.

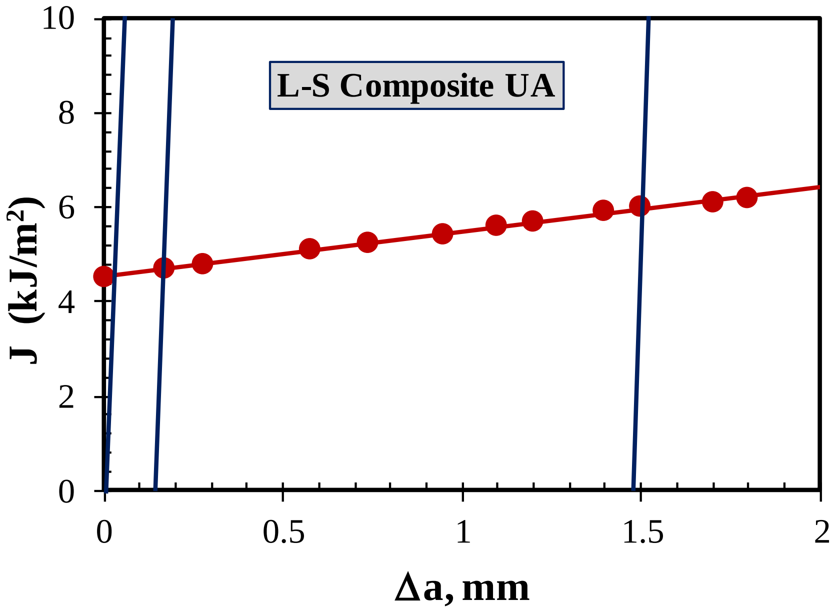

Figure 5. *J* versus Δ*a* fitted data for longitudinal metal matrix composite UA2.

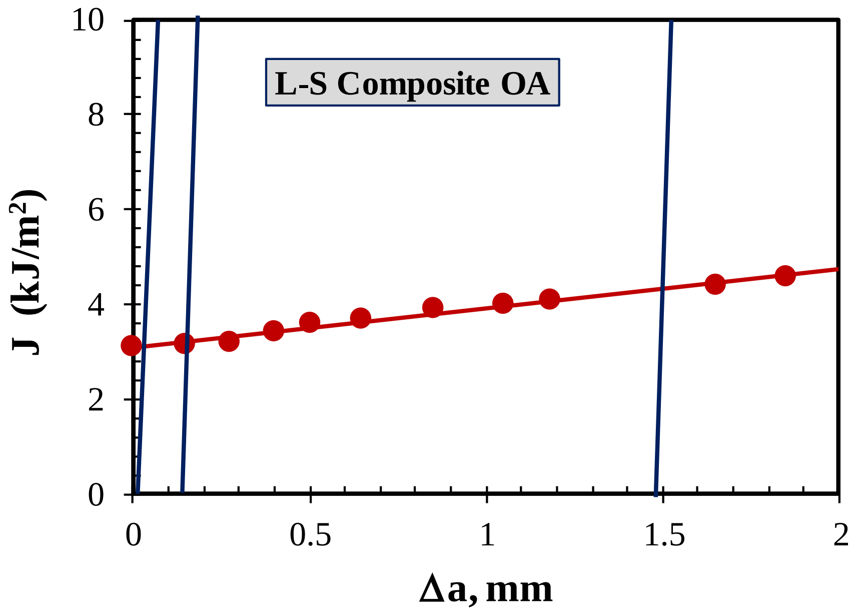

Figure 6. *J* versus Δ*a* fitted data for longitudinal metal matrix composite OA2.

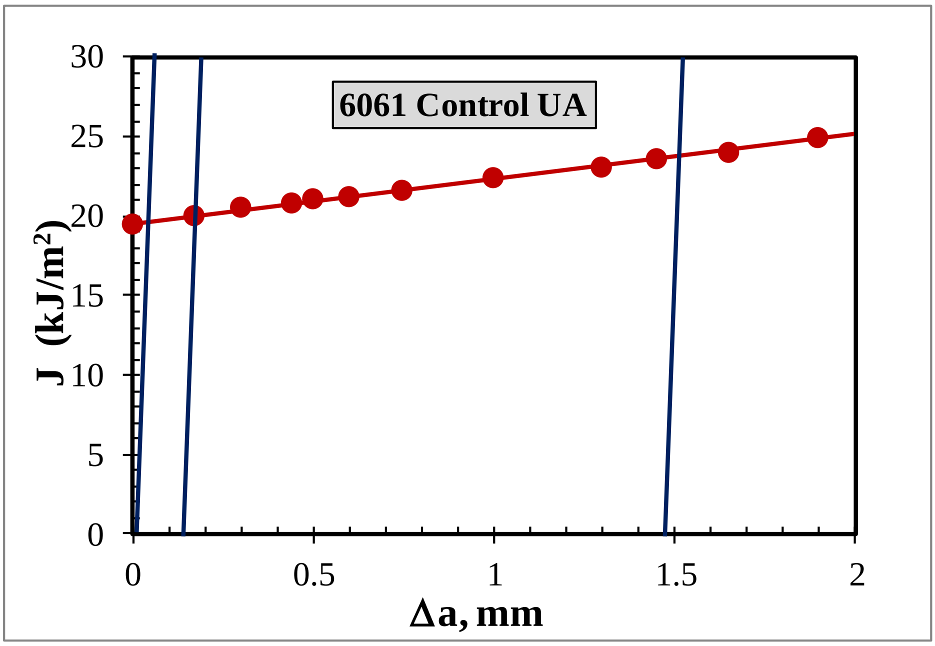

Figure 7. *J* versus Δ*a* fitted data for longitudinal control material UA.

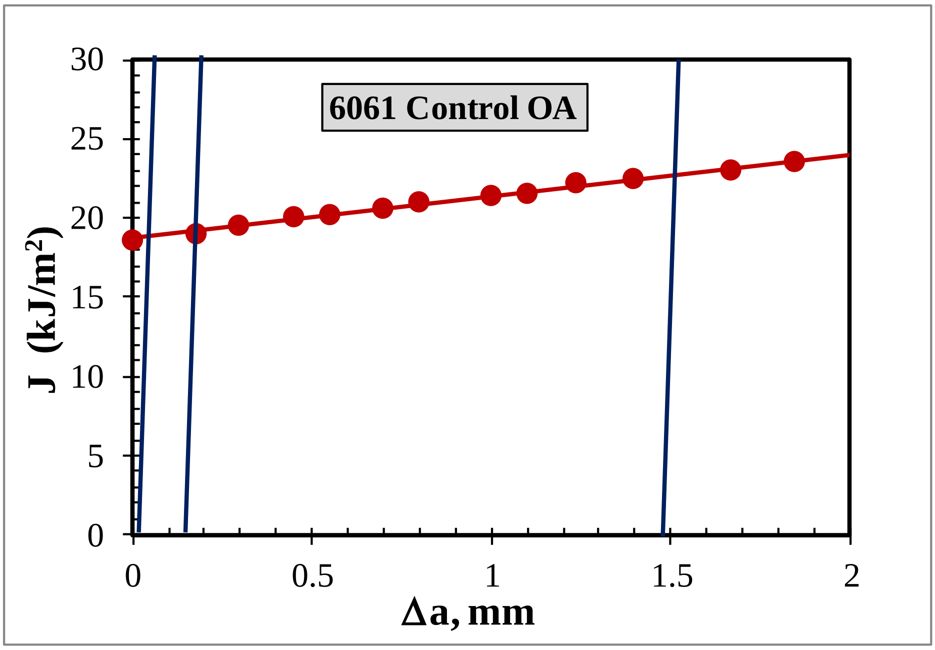

Figure 8. *J* versus Δ*a* fitted data for longitudinal control material OA.

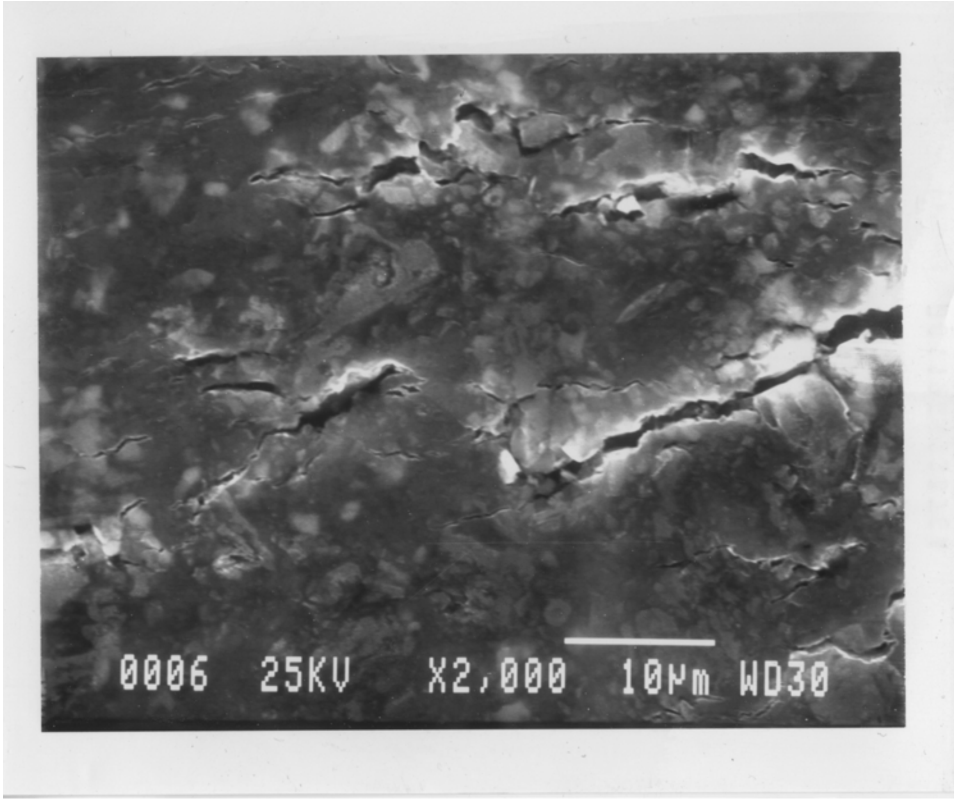

Figure 9. Crack tip region in L-S OA2 MMC showing particulate fracture, particle/matrix failure, and extensive failure in matrix at lower value of toughness (and crack tip opening) compared to UA materials.

Graphical abstract